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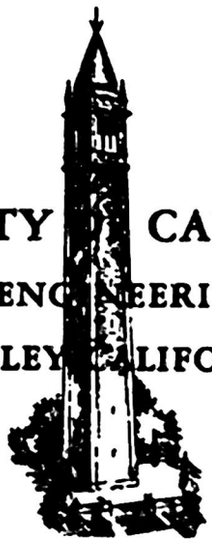

UNIVERSITY OF CALIFORNIA
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EFFECT OF STRESS ON THE CREEP RATES OF POLYCRYSTALLINE ALUMINUM ALLOYS UNDER CONSTANT STRUCTURE

Twenty Fourth Technical Report

By

O. D. Sherby, R. Frenkel, J. Nadeau and John E. Dorn

22, N7-onr-295, Task Order II,
NR-031-048
SERIES NO.....
ISSUE NO. 24
DATE February 15, 1953

March 10, 1953

Office of Naval Research
Department of the Navy
Washington 25, D.C.

ATTENTION: Dr. O. T. Marzke

Dear Sir:

Attached hereto is the Twenty Fourth Technical Report on Contract N7-onr-295, Task Order II, NR-031-048, entitled "Effect of Stress on the Creep Rates of Polycrystalline Aluminum Alloys under Constant Structure".

The wholehearted cooperation of the Office of Naval Research in making these studies possible is sincerely appreciated.

Respectfully submitted,

John E. Dorn

John E. Dorn
Professor of Physical Metallurgy

JED:bp

EFFECT OF STRESS ON THE CREEP RATES OF POLYCRYSTALLINE ALUMINUM ALLOYS UNDER CONSTANT STRUCTURE

By

O. D. Sherby [1], R. Frenkel [1], J. Nadeau [2] and John E. Dorn [3]

Twenty-Fourth Technical Report, Series 22, Issue 24,
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February 15, 1953

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ABSTRACT

A method is shown for the study of the creep rate dependence of metals on the applied stress under the condition of constant structure. The method was applied to pure aluminum and to dilute solid solution alloys of Mg, Cu, Ge, Zn, and Ag in aluminum. The results revealed that the applied stress and the creep rate are related by the equation $\dot{\epsilon} = S e^{-\Delta H/RT} \sinh B\sigma$. B was found to be independent of the creep structure for a given material; a linear relationship was found to exist between $\frac{1}{B}$ and the percent of alloying addition to aluminum for a given solute element. Furthermore $\frac{1}{B}$ is a function of the low temperature solid solution strengthening of the alloys.

INTRODUCTION

It is customary to plot the secondary creep rate as a function of the initial creep stress in order to evaluate the stress-creep rate relationship in constant load creep tests. The analyses based on such data have not been particularly rewarding insofar as no lucid correlation with reaction rate theory is achieved by this technique. The basic reasons for this failure have long been suspected and recently the failure of this approach was more clearly revealed [1].

In spite of the continuously increasing true stress over the primary stage of creep, the creep rate continuously decreases. Obviously the structural changes that occur are responsible for the observed increase in creep resistance during creep. If therefore the same structures were obtained at the secondary stage of creep independent of the applied stress, the correlation between stress and secondary creep rate would be significant for that structure. But if the structures that are developed depend on the creep stress, no unique correlation between the secondary creep rates and the initial stress can be expected from this type of analysis.

The results of recent investigations [1-4] have shown that for high temperature creep of non-precipitation hardening systems

$$\varepsilon = f(\theta, \sigma_c) \qquad (1)$$

where

$\theta = te^{-\Delta H/RT}$ = temperature-compensated time

t = time

T = absolute temperature

R = gas constant

ΔH = activation energy

ε = creep strain

σ_c = initial creep stress in a constant load test

Typical examples of this correlation are shown in Fig. 1. The validity of Equation 1 was further verified by observing that the metallographic and x-ray subgrain structures [3] are functions of the temperature-compensated time θ for a given creep stress σ_c. Additional verification [3,4] was also obtained by noting that the stress-strain curves at 298°K, which were obtained after constant load creep straining, are functions of θ alone. In this way it was demonstrated that not only the readily revealed subgrain structure, but also such currently unmeasurable structural changes (such as the number and distribution of the dislocations in the volume of the grains) are also functions of the temperature-compensated time. Thus the evidence for the validity of Equation 1 for high temperature creep is substantial.

The creep rate for a constant load creep test can now be obtained by differentiating Equation 1 with respect to time, whence

$$\dot{\varepsilon} = \left(\frac{\partial f}{\partial \theta}\right)\left(\frac{d\theta}{dt}\right) = f_\theta'(\theta, \sigma_c)\, e^{-\Delta H/RT} \qquad (2)$$

And therefore the secondary creep rate is

$$\dot{\varepsilon}_s = f_\theta'(\theta_s, \sigma_c)\, e^{-\Delta H/RT} \qquad (3)$$

But according to Equation 1 (as shown in Fig. 1) for a constant load test, θ_s becomes a function of σ_c alone, and therefore Equation 3 reduces to

$$\sigma_c = F\left(\dot{\varepsilon}_s e^{\Delta H/RT}\right) = F(Z) \qquad (4)$$

where F is some function and $Z = \dot{\varepsilon}_s e^{\Delta H/RT}$, the Zener-Holloman parameter [5]. When the experimental values of $\dot{\varepsilon}_s e^{\Delta H/RT}$ are plotted against σ_c over a range of high temperatures a single curve is obtained as shown in Fig. 2, thus verifying Equation 4. But the subgrain structures that are developed during secondary creep have been shown [3] to vary in a systematic way

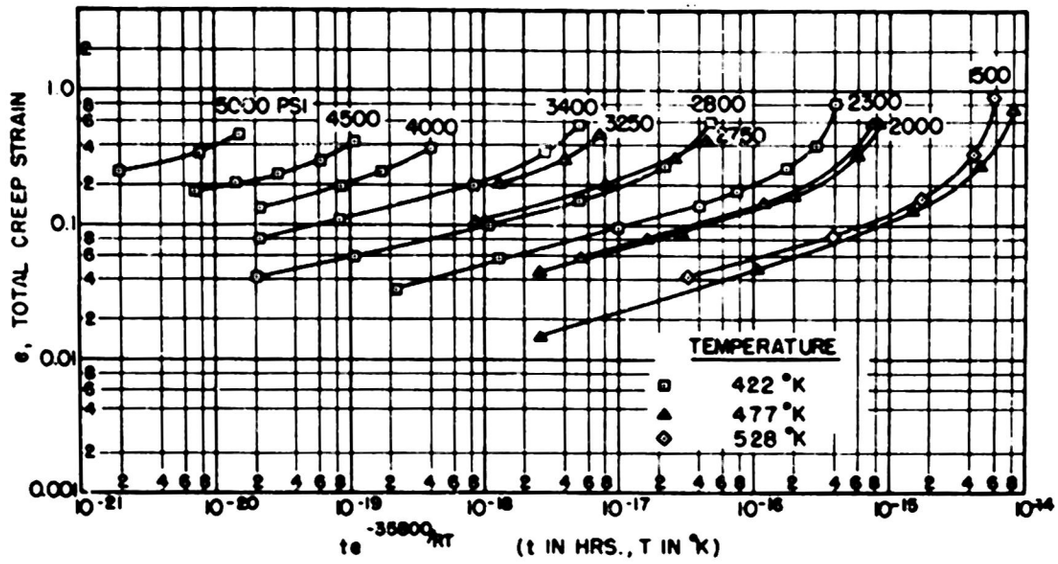

FIG. 1 CONSTANT LOAD CREEP TESTS FOR HIGH PURITY ALUMINUM. (REFERENCE 3)

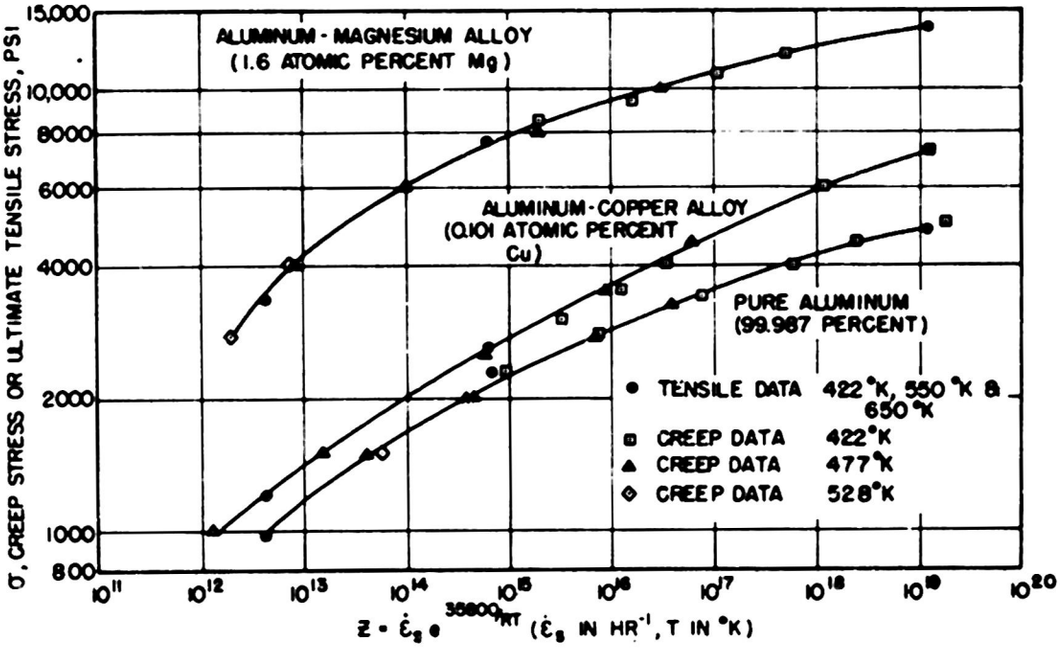

FIG. 2 CORRELATION BETWEEN THE INITIAL CREEP STRESS AND THE PARAMETER $\dot{\varepsilon}_s e^{\Delta H/RT}$ (REFERENCE 3).

with the initial creep stress, σ_c. Consequently the experimentally determined relationship between σ_c and $\dot{\varepsilon}$ is not unique because it also depends on the various structures that are produced at the secondary stage of creep under different stresses.

Fortunately a very simple technique is available for estimating the effect of stress on the creep rates of metals for a given structure. If, as previously proven, [1-4] a metal is precrept under a stress σ_1, to a strain ε_1, at any elevated temperature a definite structure will be produced. An introduction of a higher stress at this point will cause a rapid change in structure due to the additional instantaneous plastic strain imposed on the metal. But if the stress be reduced to a value below σ_1, after precreep to ε_1, the structure of the metal might be expected to initially remain that of the precrept state. For relatively short times at the lower stresses, therefore, the structure will be expected to differ only slightly from that obtained at the end of the precreep treatment, the changes occurring being due to a small amount of creep recovery, some crystal recovery, and those changes in structure attending the creep at the lower stress. Assuming that these changes are small and thus have a minor effect on the creep rate, it is then possible to estimate the effect of stress on the creep rate for a constant structure by a series of tests in which the initial creep stress σ_1 is reduced to a series of lower stress values following precreep to a given strain ε_1. Auxiliary tests, to be reported at a later date, have shown that the amount of creep recovery attending complete unloading of the alloys used in this investigation is indeed negligible and therefore this factor does not contribute in any material way to possible changes in structure upon reducing the creep stress. Furthermore any change in structure due to crystal recovery or creep straining at the new stress should be reflected in appropriate changes of the creep rate with strain. Inasmuch as the strain-time curves obtained in the investigations to be

outlined in this report were reasonably linear over small strains following a reduction in stress, it seemed appropriate to assume that crystal recovery and the auxiliary structural changes attending the small additional creep had at most a negligible effect on the interpretation of the data.

MATERIALS AND TECHNIQUES

The solid solution aluminum alloys that are listed in Table I were used in the present investigation. Sheets of these alloys were homogenized, cold rolled from 0.100 to 0.070 inches in thickness and then recrystallized to about the same grain size. The creep specimens were selected with their tensile axes in the rolling direction. All creep tests were conducted under constant load conditions. The strain was measured to $\pm$ 0.0001 and the initial creep stress was measured to $\pm$ 20 psi.

RESULTS AND DISCUSSION

A. High Purity Aluminum

A typical example of one series of tests on high purity aluminum is shown in Fig. 3. Each specimen was precrept to an engineering strain of 25% at an initial stress of σ_c = 3400 psi. Small differences in creep rates due to unknown sampling problems were noted in the creep curves for the various specimens. When an engineering strain of 25% was reached the load was reduced to give new reduced engineering stresses of 3,200 psi etc. to 2,500 psi as shown in the figure. Although difficulties were anticipated in determining the instantaneous creep rate immediately after reducing the stress, all of the creep curves exhibited rather good straight lines over the initial intervals of creep following reduction of the stress. Consequently fairly reliable initial creep rates could be obtained for the various reduced stresses.

The data shown in Figure 3 were analyzed in terms of true stresses and true

TABLE I

Chemical Analyses and Grain Size of Alloys

Alloying Element	Atomic Percent	Grain Size Diam. mm.	Chemical Analyses (wt. % impurities) Si	Fe	Cu	Mg	Mn
Aluminum (99.987)		0.21	.003	.003	.006	.001	
Magnesium	0.554	0.25	.003	.003	.007		
	1.097	0.28	.004	.004	.007		
	1.617	0.26	.003	.004	.006		
Copper	0.101	0.29	.003	.003		.0006	.001
	0.232	0.30	.003	.004		.0006	.001
Zinc	0.755	0.26	.004	.005	.006	.001	
	1.616	0.26	.003	.005	.007	.001	
Germanium	0.082	0.27	.003	.005	.007	.001	
	0.145	0.26	.003	.006	.007	.001	
Silver	0.100	0.29	.003	.005	.007	.001	
	0.194	0.29	.003	.006	.007	.001	

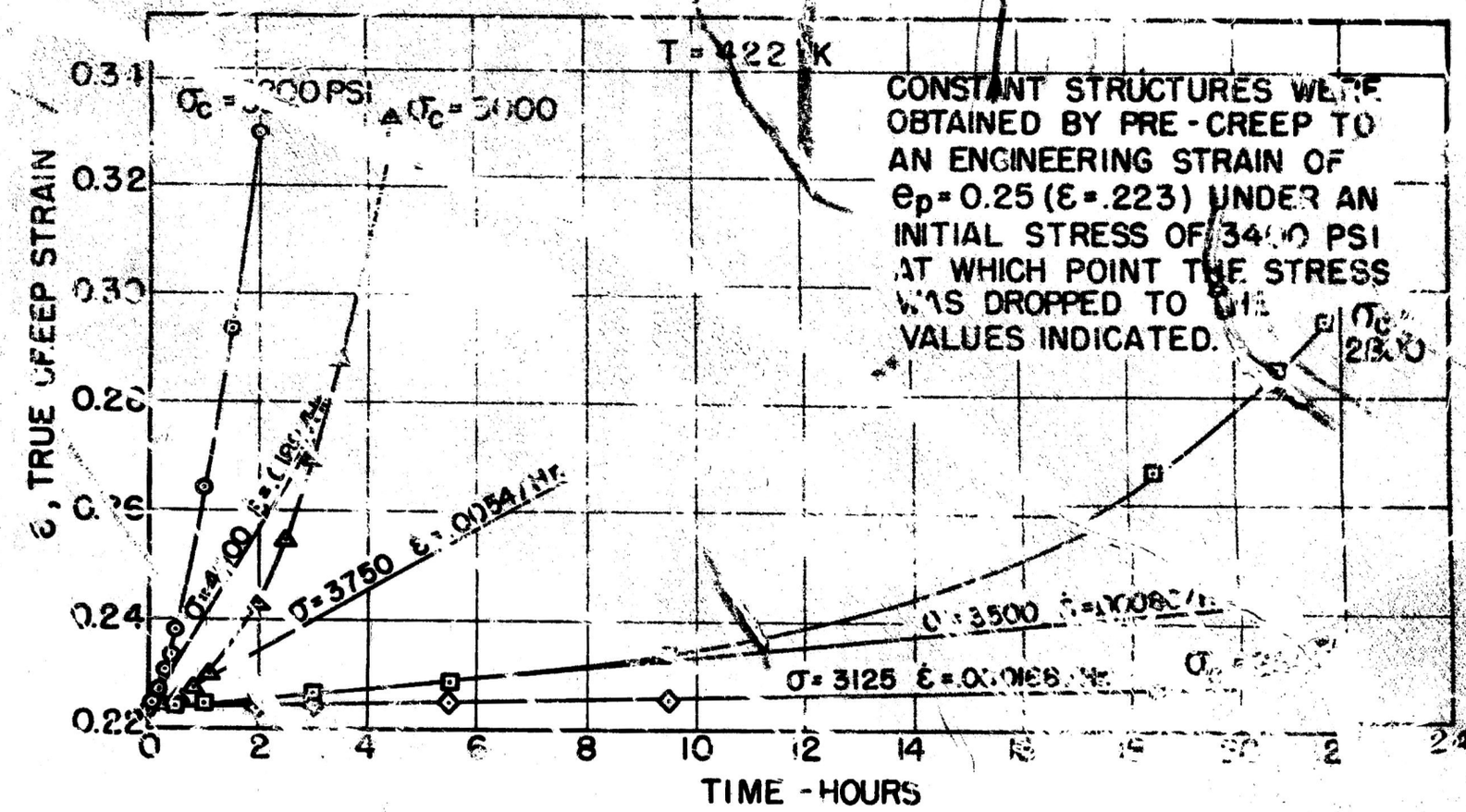

FIG 3 TYPICAL CREEP CURVES FOR HIGH PURITY ALUMINUM [illegible] CREEP RATE DEPENDENCE ON THE STRESS UNDER CONSTANT STRUCTURE CONDITIONS.

plastic strain rates according to the following:

$$\sigma = \frac{\text{Load}}{\text{Instantaneous Area}} = \frac{L}{A} = \frac{L}{A_o} \cdot \frac{A_o}{A} = \sigma_c \left(\frac{l}{l_o}\right) = \sigma_c (1+e_p) \qquad (5)$$

where e_p is the engineering plastic strain of the test, σ_c is the engineering stress, and

$$\dot{\varepsilon} = \frac{d\varepsilon}{dt} = \frac{d\left(\ln \frac{l}{l_o}\right)}{dt} = \frac{d\left\{\ln (1+e_p)\right\}}{dt} \qquad (6)$$

where $\dot{\varepsilon}$ is the instantaneous true strain rate after unloading to the lower stress level.

Since the precreep conditions were the same, the σ-$\dot{\varepsilon}$ relationship refers to a given structural state. This type of correlation was attempted for other structural states developed by various precreep conditions; creep tests were performed at initial creep stresses of 3400 psi and 422°K to strains of e_p = 9% and 15% to obtain different creep structures at which point the stresses were again decreased appropriately and the relationship between the instantaneous true stress and instantaneous true creep rate was determined. Furthermore, another temperature of test, 530°K, to strains of e_p = 9% and 25% under an engineering stress of σ_c = 2000 psi was also used. The creep rate dependence on stress for these various structures is illustrated in Fig. 4. Several significant deductions are immediately apparent from these data.

1. The true creep rate-true stress relationship for a constant structure is given by

$$\dot{\varepsilon} = S' 10^{B'\sigma} = S'' e^{B\sigma} \qquad (7)$$

where $\frac{1}{B}$ is merely the slope of the linear σ versus log $\dot{\varepsilon}$ curves of Fig. 4 divided by the constant, 2.303. Since the creep rate must vanish when the stress σ is reduced to zero, it appears that equation 7 can only be valid at high stresses and therefore the general relationship is probably

$$\dot{\varepsilon} = S''' \sinh B\sigma \qquad (8)$$

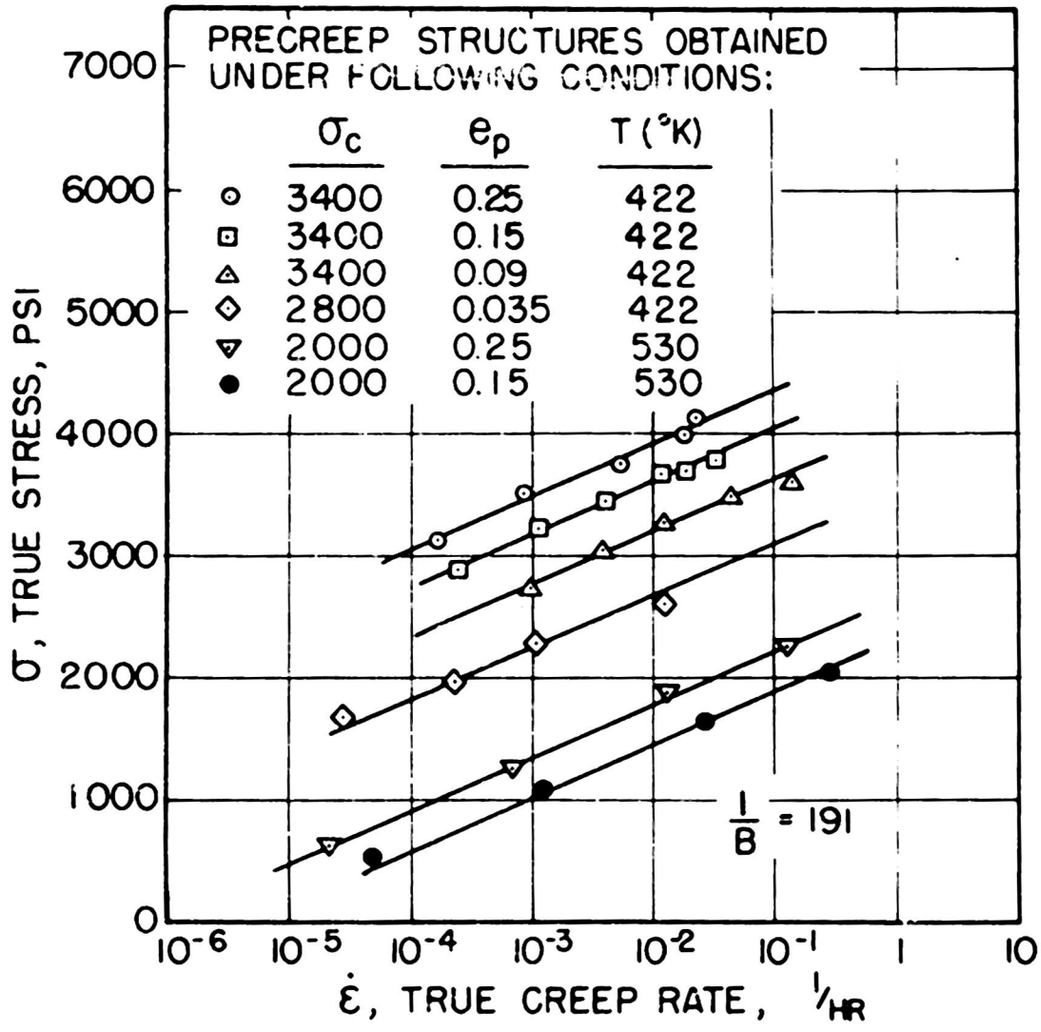

FIG. 4 EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR HIGH PURITY ALUMINUM.

2. Furthermore the data of Fig. 4 reveal in two separate ways that the value of B is independent of the structural changes induced by the precreep treatments. In previous investigations [3] it was shown that the structures of aluminum crept to the same strains under different stresses were substantially different. But since the values of B deduced from Fig. 4 are independent of the precreep stress, σ_c, B must be independent of the structure. In addition it is now well documented [3] that the structure of aluminum changes greatly with creep strain for a given creep stress. Consequently the observations that B is a constant over the ranges from 9.0% to 25% precreep strain prove conclusively that B is independent of structure.

3. Formal applications of reaction rate theory to the creep of metals suggest [6] that the creep rate should be approximated by the relationship

$$\dot{\epsilon} = \alpha e^{-\Delta H/RT} \sinh \frac{\beta \sigma}{T} \tag{9}$$

where β is related to the cross-sectional area of a dislocation times the distance it must be moved under stress in order to be activated. But the data given in Fig. 4 show that σ and not $\frac{\sigma}{T}$ is linearly related to the logarithm of the strain rate. Two additional proofs that the creep phenomenon depends on σ and not $\frac{\sigma}{T}$ have already been presented in the two sets of investigations leading to the formulation of Equations 1 and 3 respectively. Since the assumption that β might increase linearly with the absolute temperature is wholely untenable, the combinations of Equations 2 and 8 demand that the creep strain rate be given by

$$\dot{\epsilon} = S e^{-\Delta H/RT} \sinh B\sigma \tag{10}$$

where now both ΔH and B are known to be independent of the structure that is developed during creep; therefore all of the structural changes that occur during creep are incorporated in the parameter S.

Several major factors must be resolved before a better understanding of the phenomenon of creep can be formulated. Perhaps the most significant of these is the unexpected result that the hyperbolic sine term contains not $\frac{\sigma}{T}$ but only σ. This fact arises from the basic details of the activation process for creep. The second factor is associated with the S parameter. Although from a purely phenomenological viewpoint, a comparison of Equations 2 and 10 reveals that S is merely a function of the temperature-compensated time (or the strain) and the stress, σ_c, the parameter S must eventually be correlated with the significant structures that determine the entropy, frequency and shear strain per unit activation. Furthermore it is now known that Equation 10 is valid only above about 0.45 of the melting temperature [2,3,7] whereas substantial creep can yet take place at lower temperatures. Perhaps additional complicating factors enter the low temperature creep phenomenon.

B. Effect of Alloying Elements

The effect of various alloying elements on B was evaluated by the procedures previously described for pure aluminum for the various alloys identified in Table I. All original data on these alloys are recorded in the graphs of the appendix. The data of Fig. 5 reveal that $\frac{1}{B}$ increases almost linearly with atomic percent of the solute element. Furthermore those elements that were previously shown to have the greatest effect on solid solution strengthening at low temperatures [8] also appear to exhibit the greatest effect on inhibiting creep insofar as they exhibit the greatest value of $\frac{1}{B}$. The excellent correlation between the tensile deformation strength at 5 percent strain at 194°K and $\frac{1}{B}$ for creep is shown in Fig. 6. Perhaps this correlation arises from the fact that $\frac{1}{B}$ as well as low temperature solid solution strengthening are both dependent on the strain-energy interactions and electronic interactions between solute atoms and dislocations. The total effect of solute atoms on creep, however, is not neces-

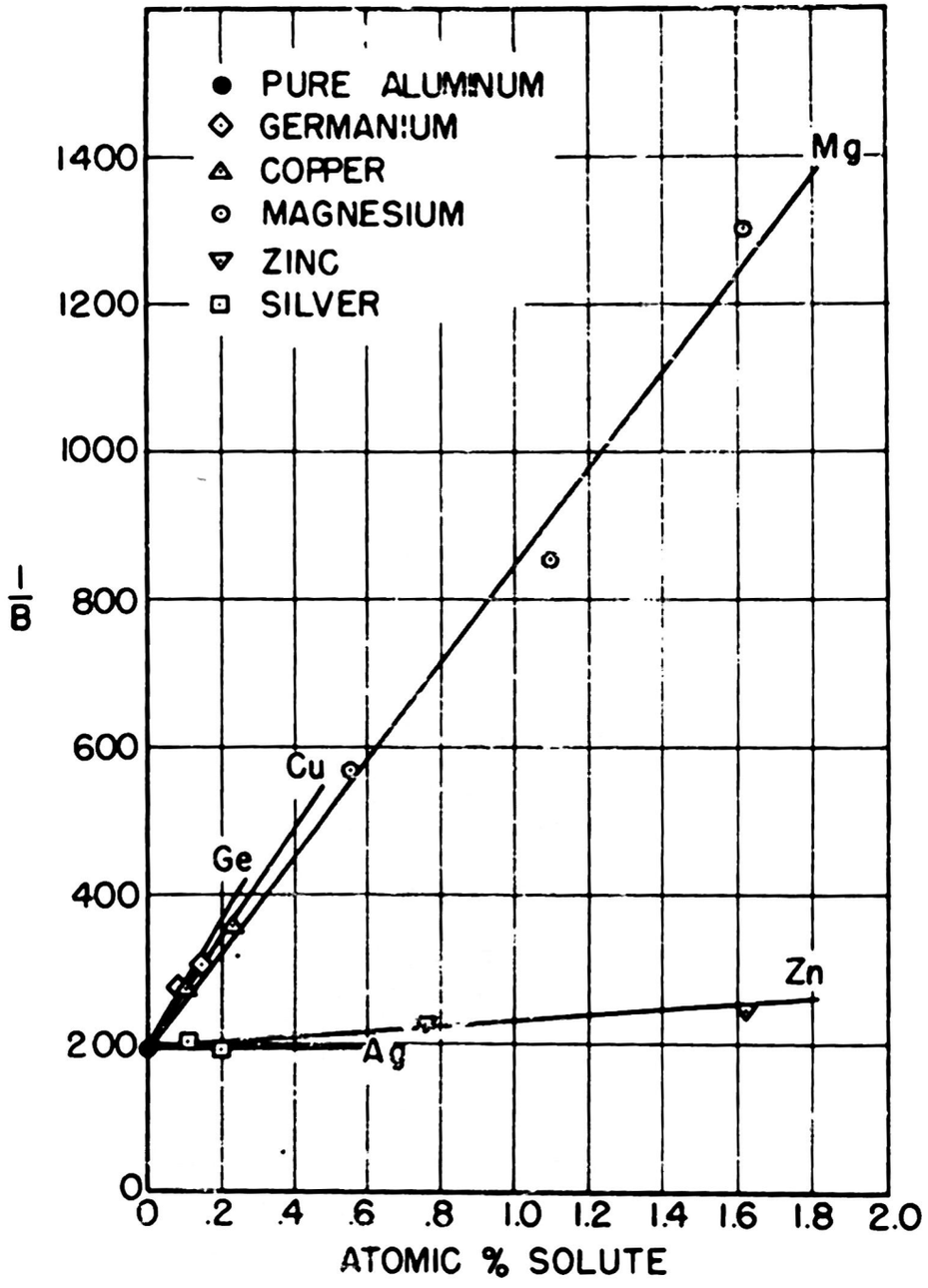

FIG. 5 EFFECT OF ALLOYING ELEMENTS ON THE PARAMETER $\frac{1}{B}$ FOR CREEP.

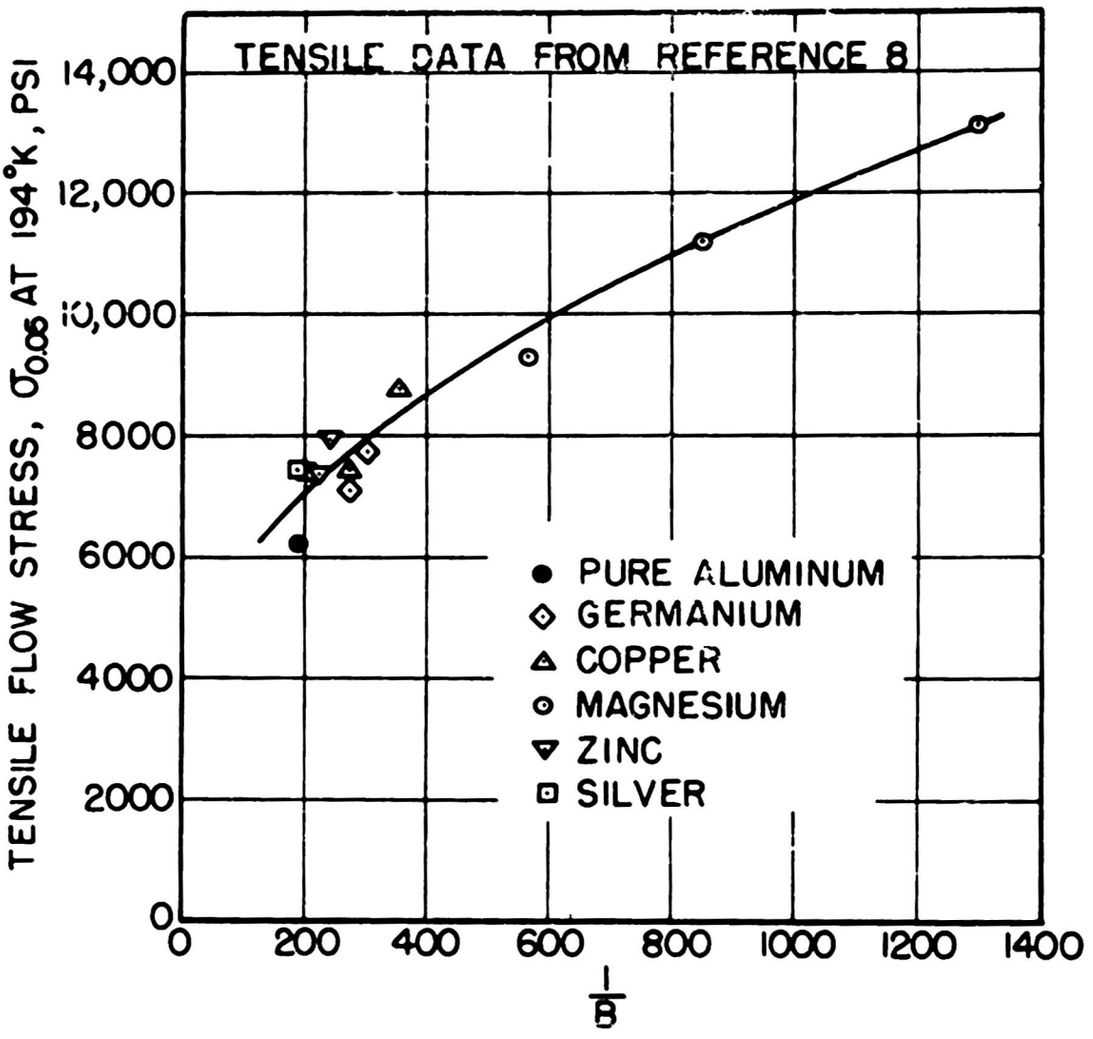

FIG. 6 CORRELATION BETWEEN $\frac{1}{B}$ FOR THE VARIOUS SOLID SOLUTION ALLOYS AND THE FLOW STRENGTH AT 194 °K.

sarily revealed by the $\frac{1}{B}$ values; solute atoms might modify the deformation structure and restrain the necessary processes that attend creep so as to change S .

CONCLUSIONS

1. A method was developed for determining the effect of stress on the creep rate of metals at constant structures.

2. The equation for the creep rate was found to be

$$\dot{\epsilon} = S e^{-\Delta H/RT} \sinh B\sigma$$

where ΔH and B are independent of the creep structure and S is a parameter that depends on structure.

3. The value of $\frac{1}{B}$ increases almost linearly with atomic percent of solute atoms. Furthermore $\frac{1}{B}$ is a function of low temperature solid solution strengthening.

ACKNOWLEDGMENTS

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APPENDIX

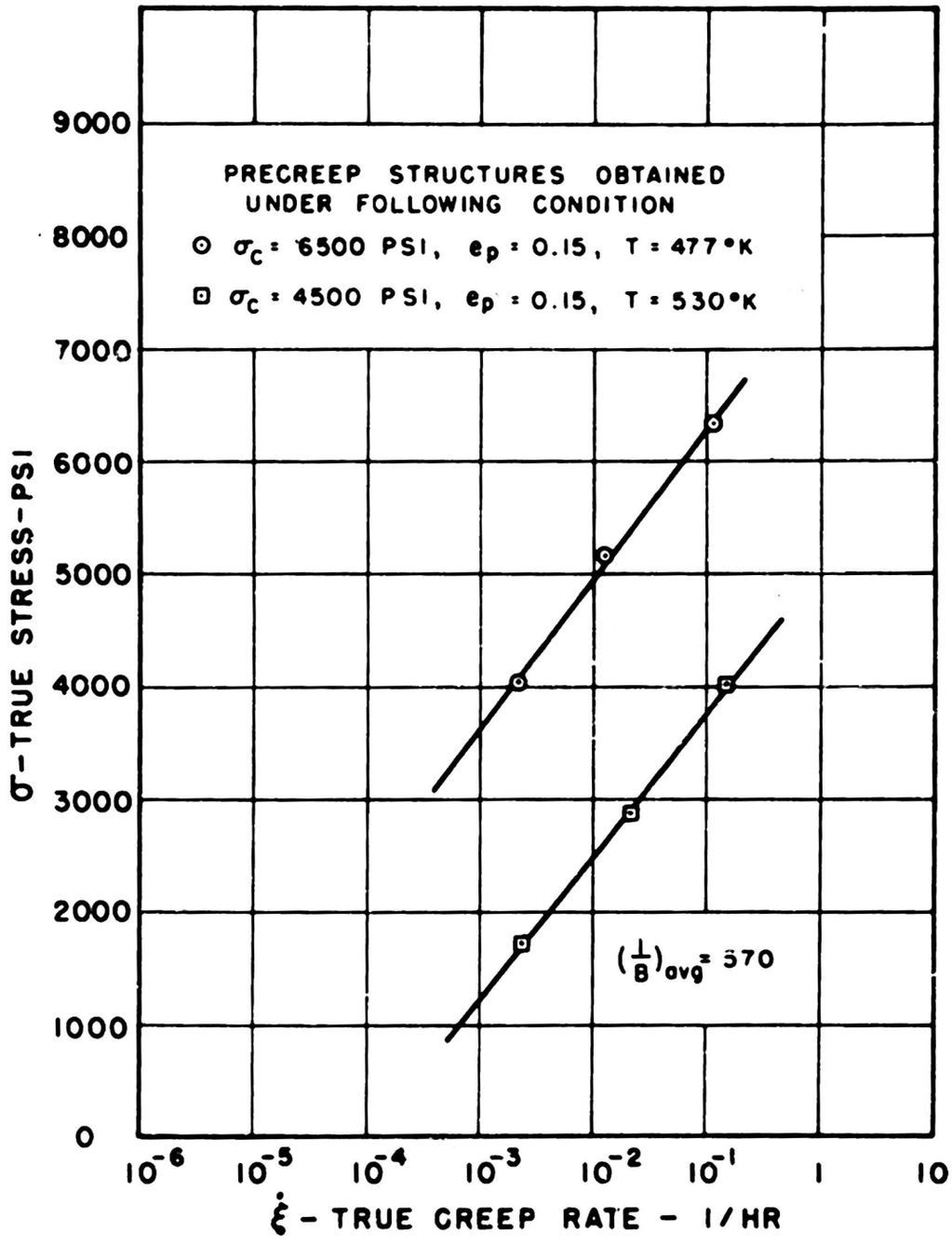

FIGURE A1. EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR 0.5 ATOMIC PERCENT MAGNESIUM IN ALUMINUM SOLID SOLUTION ALLOY.

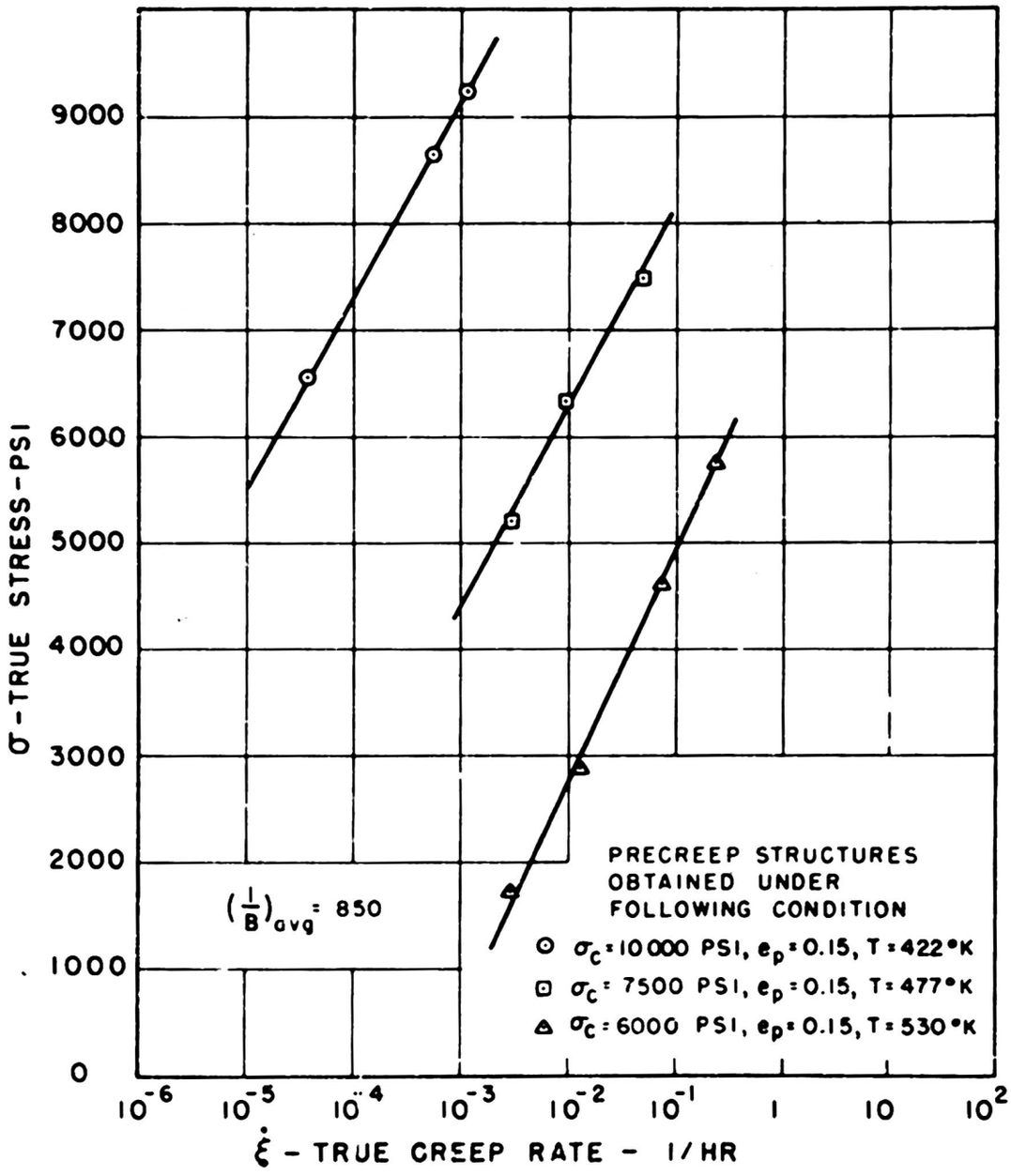

FIGURE A2. EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR 1.10 ATOMIC PERCENT MAGNESIUM IN ALUMINUM SOLID SOLUTION ALLOY.

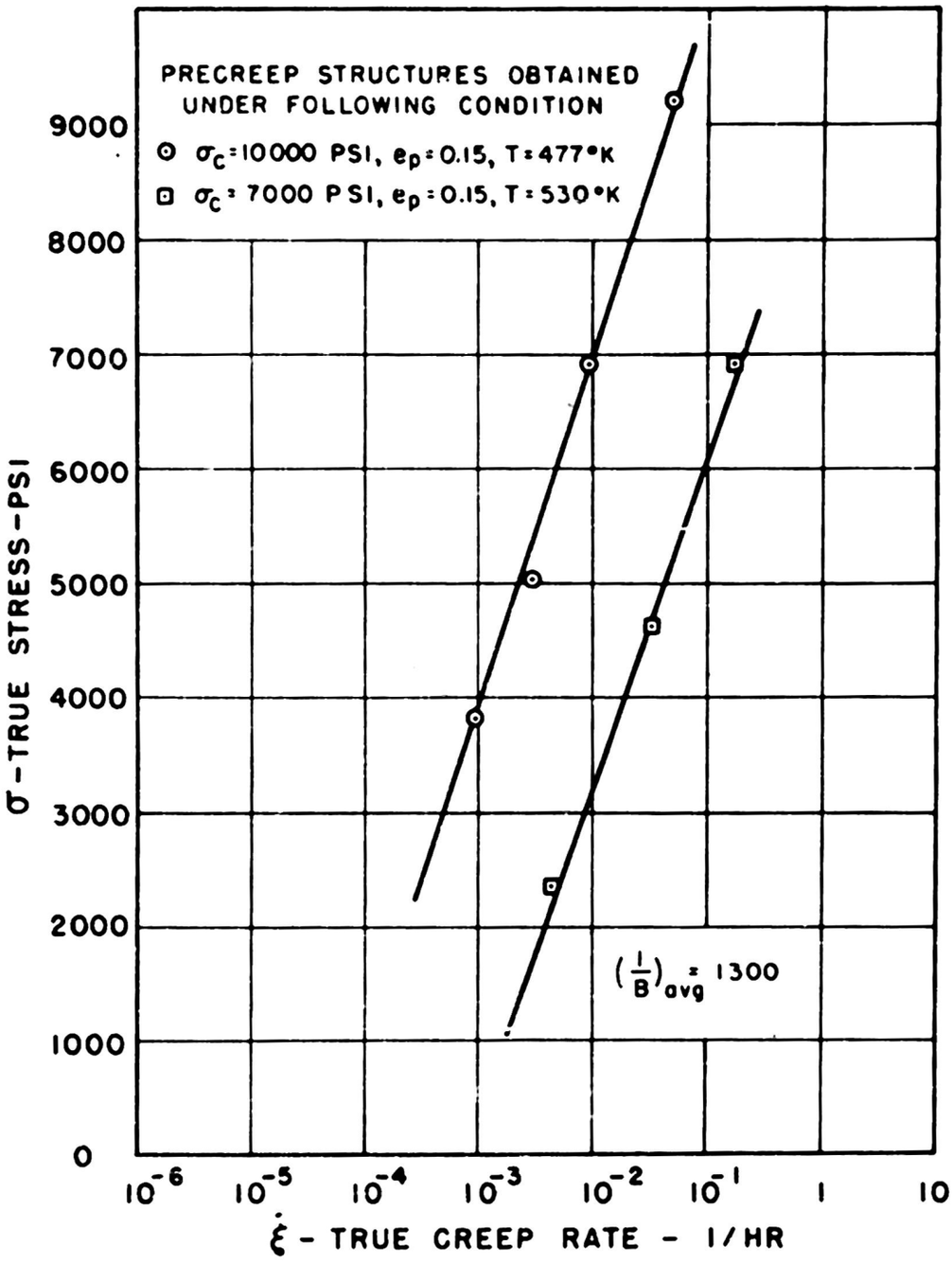

FIGURE A3. EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR 1.62 ATOMIC PERCENT MAGNESIUM IN ALUMINUM SOLID SOLUTION ALLOY.

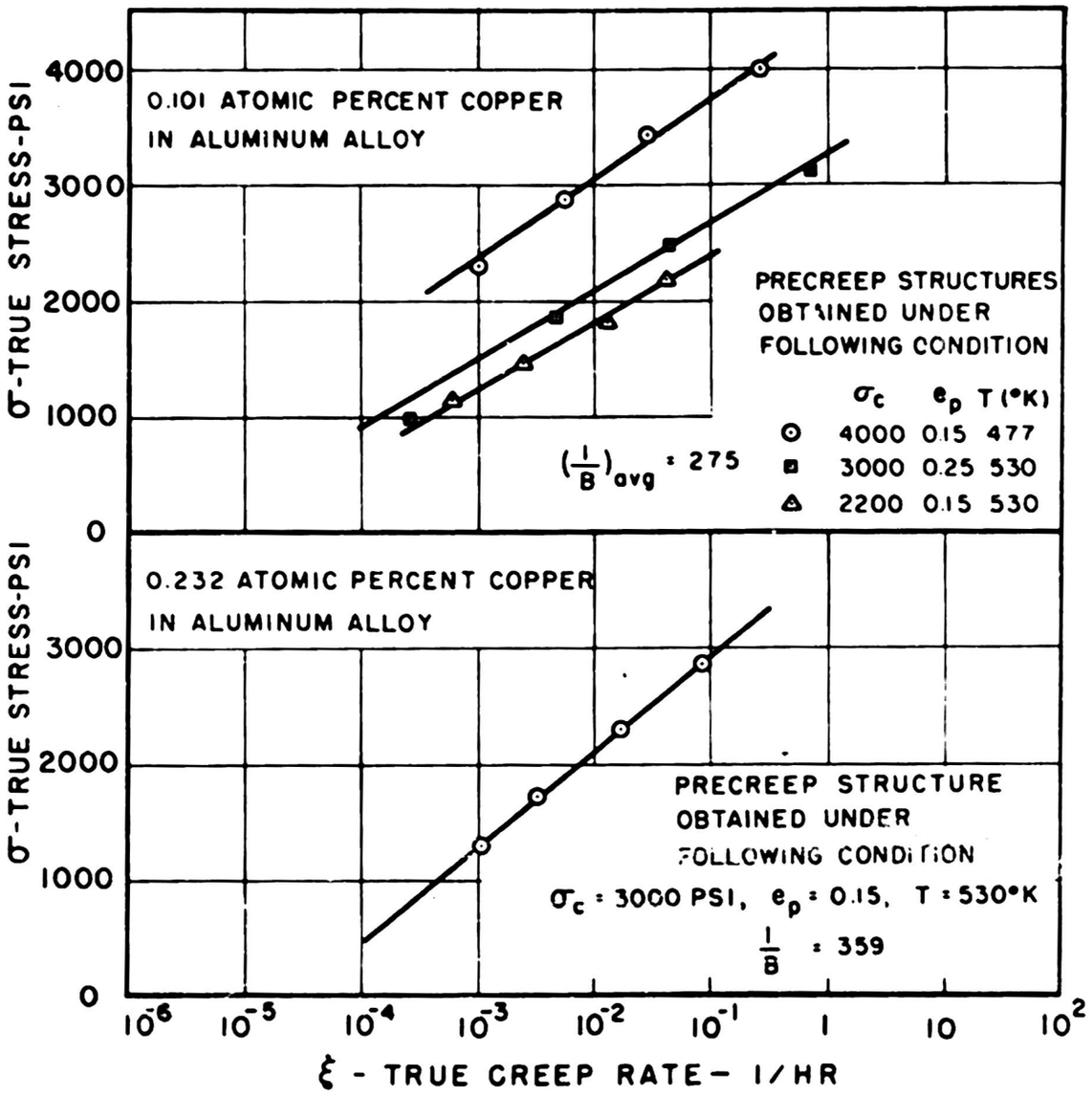

FIGURE A 4. EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR Al-Cu SOLID SOLUTION ALLOYS

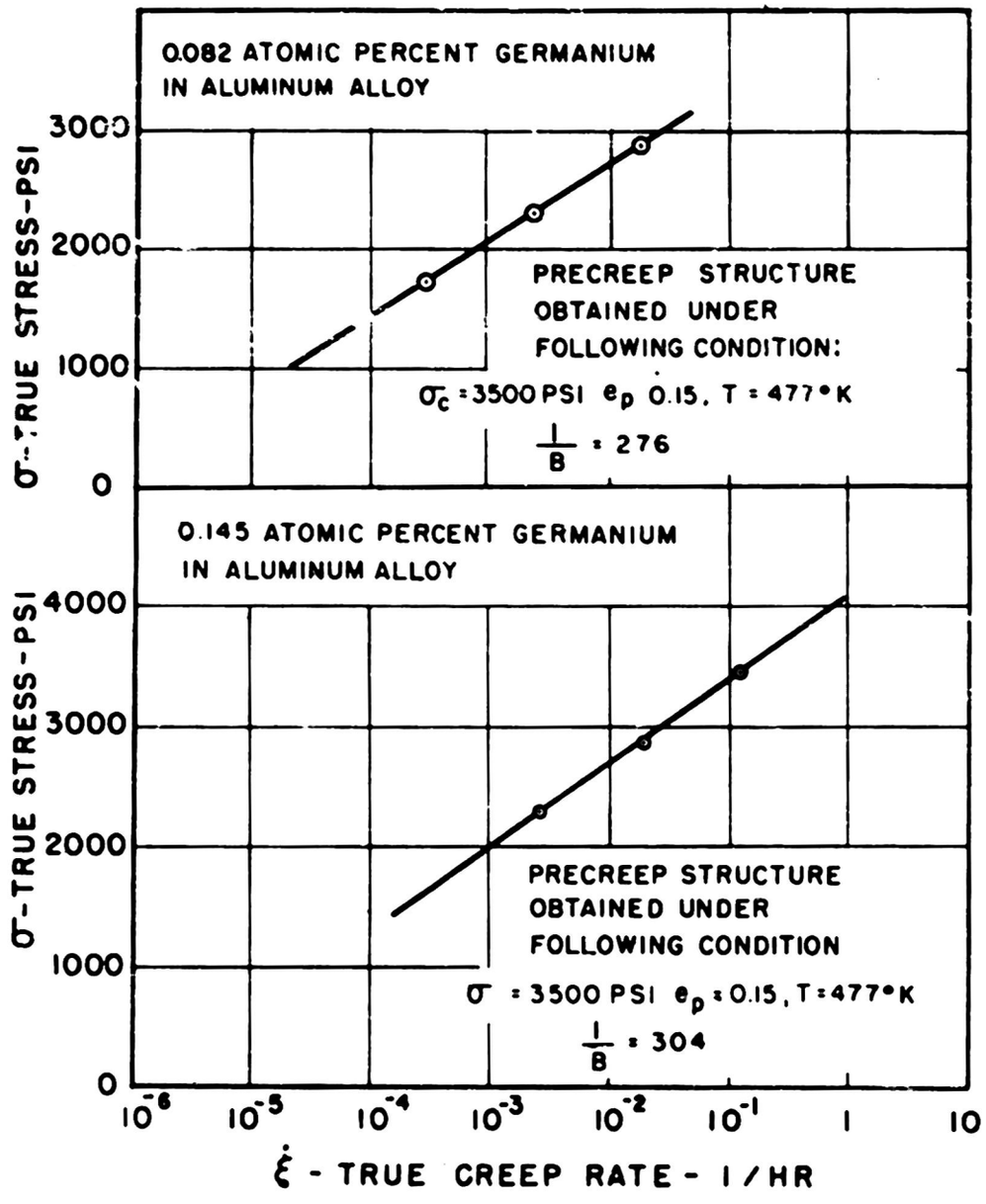

FIGURE A5. EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR Al-Ge SOLID SOLUTION ALLOYS

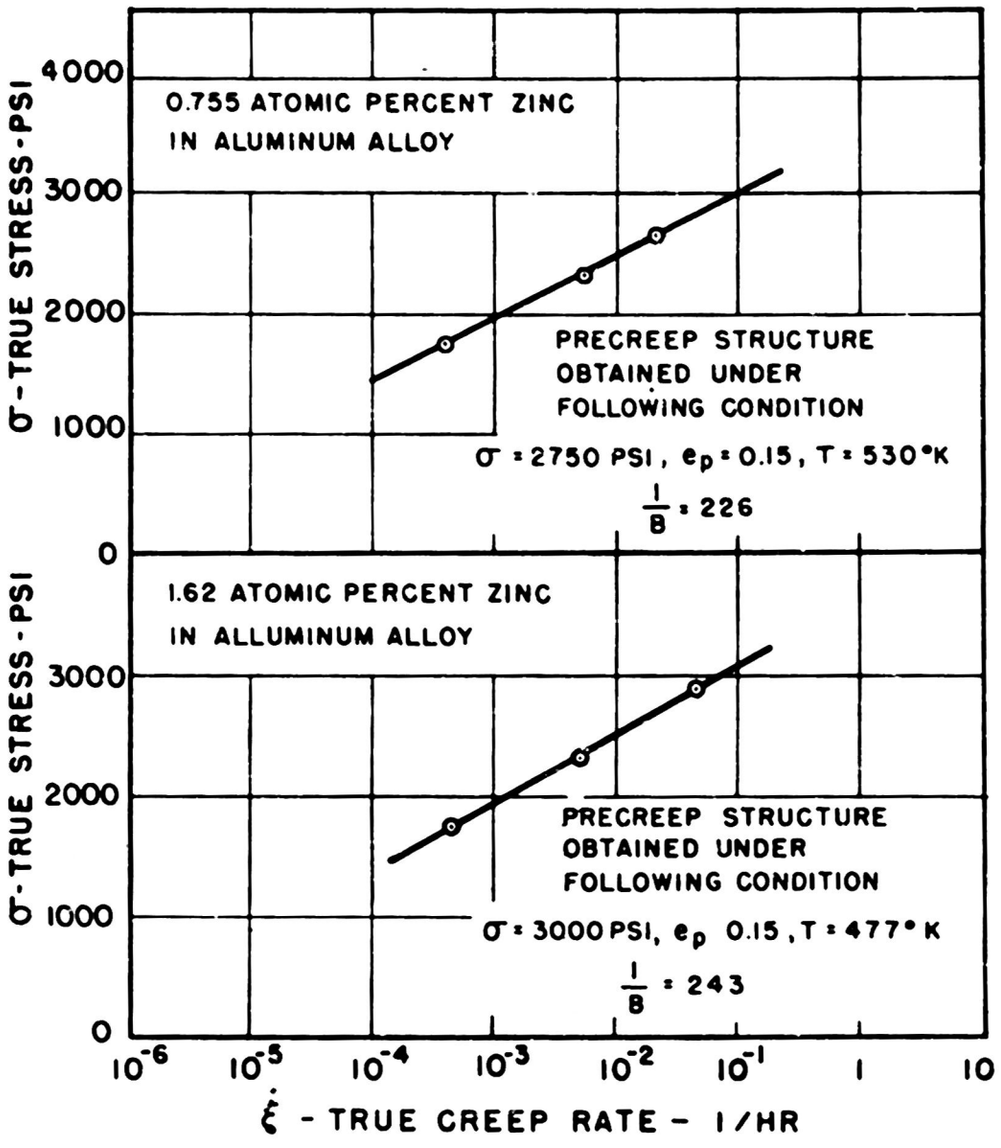

FIGURE A6. EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR Al-Zn SOLID SOLUTION ALLOYS

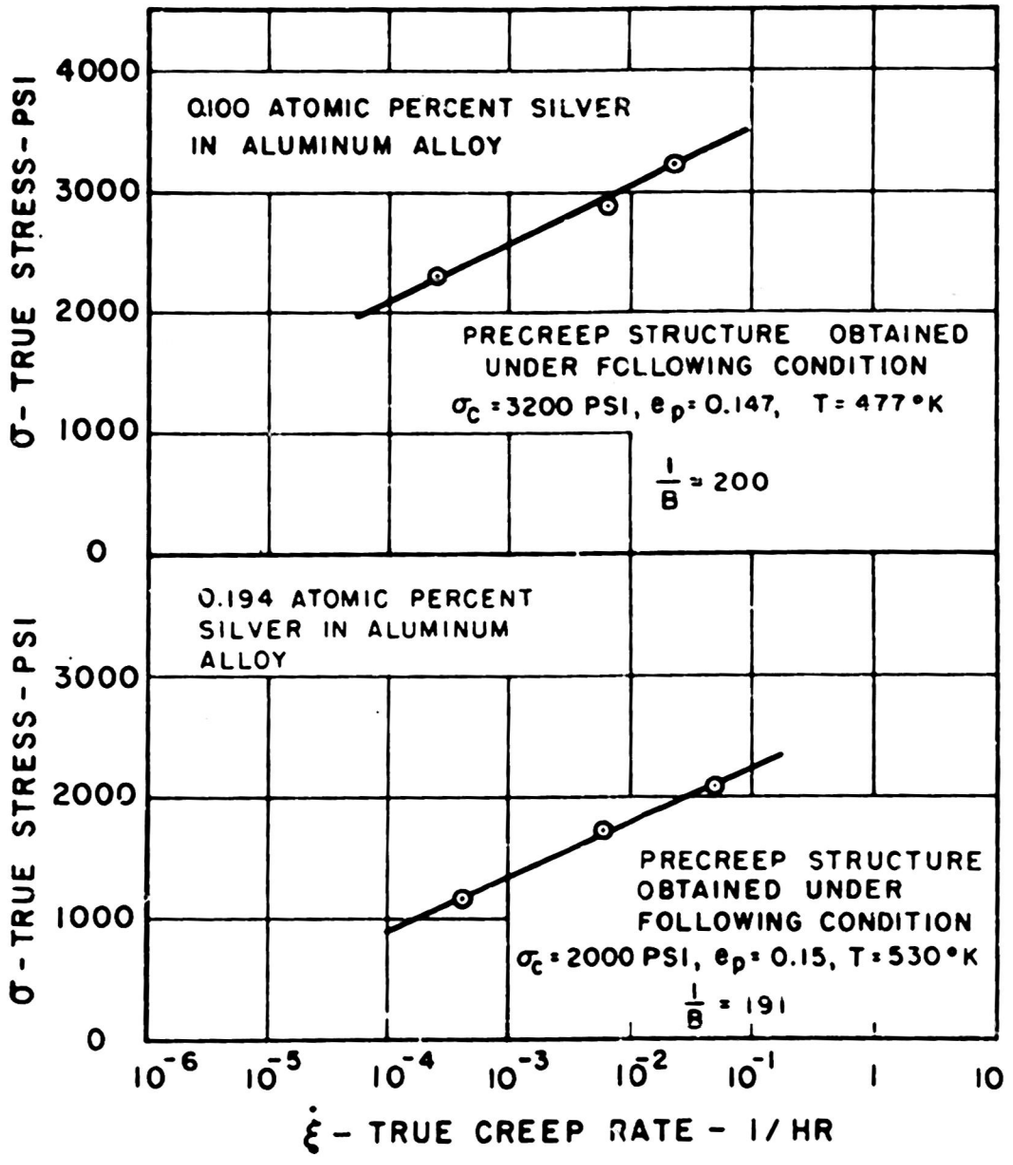

FIGURE A7. EFFECT OF TRUE STRESS ON THE TRUE CREEP RATE AT CONSTANT STRUCTURE FOR Al - Ag SOLID SOLUTION ALLOYS.

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